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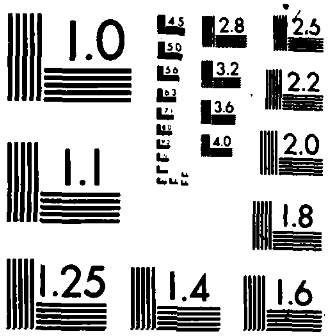

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Dr. Hans Conrad 18184-MS
North Carolina State University
Department of Materials Engineering
Raleigh, NC 27650

THE ELECTROPLASTIC EFFECT IN METALS*

H. Conrad, A. F. Sprecher and S. L. Mannan**
Materials Engineering Department
North Carolina State University
Raleigh, NC 27695-7907

ABSTRACT

The influence of electric current on the mechanical properties of metals is reviewed. Included are the effects of both low density ($\sim$1A/mm^2) continuous current and high density ($\sim$10^{3}A/mm^2) current pulses, with special attention given to the latter. The influence of the high density current pulses is analyzed in terms of the parameters involved in the thermally activated overcoming of short-range obstacles to dislocation motion. Considered are: (a) the force drift electrons exert on dislocations held up at obstacles and (b) the influence of drift electrons on the obstacle strength and on the pre-exponential factor.

THAT THE DRIFT of free electrons in a metal crystal may interact with the dislocations therein was first reported by Troitskii and Lichtman in 1963(1). They found during electron irradiation of Zn single crystals undergoing plastic deformation that there occurred a significant decrease in the flow stress and an improvement in ductility when the electron beam was directed along the (0001) slip plane, compared to when it was directed normal to the plane. Similar results were subsequently reported by Troitskii in 1968(2). These observations led to the conclusion that drift electrons exert a force ("electron wind") on dislocations and therefore such force should also occur during the passage of an electric current(3,4). Theoretical support for a drift electron-dislocation interaction is provided by the considerations of Kravchenko (5), Klimov, Shnyrev and Novikov(6), and more recently Roshchupkin, Miloshenko and Kalinin(7).

The idea that drift electrons can influence the generation and motion of dislocations led Troitskii and other Russian scientists to carry out a series of investigations into the influence of direct current pulses of the order of 10^{3}A/mm^2 for 100μs on the mechanical properties of metals at room temperature and below, including the flow stress(3,[illegible]-11), stress relaxation(12-15), creep (16,17), dislocation generation and mobility(18, 19), brittle fracture(20-22), fatigue(23), and metal working(16,24-33). The observed effect of electric current pulses on mechanical properties was termed an electroplastic effect.

Papers by North American workers on the electroplastic effect in metals began appearing in the literature from 1979(34-44). Shortly thereafter papers also began to appear from other regions of the world. In 1980, San Martin and coworkers(45)of East Germany reported an electroplastic effect in the intermetallic compound V_3Si deformed at elevated temperatures, indicating that the effect is not restricted to metals deformed at relatively low temperatures. In 1981, Silveira and coworkers(46,47) of Brazil found that the rate of stress relaxation of a number of metals at elevated temperatures increased significantly with the application of a small, continuous direct current of 1.6A/mm^2, indicating that the influence of an electric current on the mechanical properties of metals is not restricted to high current densities. This was in keeping with an earlier observation of Kishkin and Klypin(48), who found that the creep rate of a number of metals increased upon application of currents as low as 0.15A/mm^2.

In addition to an influence of drift electrons on the plastic flow stress or rate, changes in the dislocation structure have been observed (49,50); also changes in texture and in subsequent tensile properties have resulted from the application of current pulses during metalworking (32,51,52). Further, it was found(53,54) that the concurrent application of high density d.c. current pulses during the annealing of copper increased the rates of recovery, recrystallization and grain growth, and reduced the recrystallized grain size and number of annealing twins per grain. A decrease in the recrystallized grain size of copper has also been noted by Silveira et al.(55) for relatively smaller, continuous d.c. current densities of 6.5 to 15.5A/mm^2. Of related interest are the observations that electric currents have been found to influence the

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** On leave from Reactor Research Centre Kalpakkam, Tamil Nadu, India

rates of precipitation in Fe-C(56,57) and in Al-4wt.% Cu(58,59) alloys. Also worthy of mention is that the heating of Cu and Sn/Pb layers on Al wires by passage of an alternating electric current of $10A/mm^2$ led to the formation of η' and ϵ intermetallic phases, whereas only η' formed when the specimens were heated indirectly(60). Additional observations on the effects of electric fields and/or currents on the structure and properties of metals are presented in Refs. 48, 61-64.

The present paper will review the work on the influence of an electric current on the mechanical properties of metals, giving attention to: (a) separating out the side effects of the current, such as, for example, Joule heating, the pinch effect and the skin effect, and (b) identifying the role of electron-crystal defect interactions.

The term electroplastic effect has been used by some (e.g., Troitskii and coworkers) for the total effect of the current (including the side effects) and by others (e.g., Conrad and coworkers) for the difference between the total effect and the side effects. The difference in use of the term resulted from the fact that in the early work by Troitskii on Zn single crystals the side effects were relatively small compared to the total effect, so that the total effect was considered to be due primarily to an electron-dislocation interaction. In the more recent work by Troitskii, side effects of the current (especially Joule heating) have been found to represent a significant fraction of the total effect. In view of this, Troitskii coined the term electronoplastic effect for the component due to electron-dislocation interactions, and retained the term electroplastic for the total effect of the electric current. For the sake of consistency and to avoid confusion, the present paper will follow the terminology proposed by Troitskii that the electroplastic (ep) effect refers to the total effect of the electric current. However, we will employ the term electron-crystal defect (ecd) effect for the portion due to the interaction of drift electrons with crystal defects in general. The term electron-dislocation (ed) effect will then refer to the interaction of electron with dislocations.

In view of differences in objectives and approaches by each of the various regional groups of investigators listed above, the plan to be followed in the present paper will be to first review the results of each group separately and then, in a more general discussion, attempt to draw from these results the present state of our knowledge and understanding of the subject.

RUSSIAN WORK

FLOW STRESS

General - Most of the Russian work on the effects of electric current on the flow stress of metals has been by Troitskii and coworkers, early reviews of which are found in Refs. 9 and 10. Their studies were mainly on Zn single crystals tested in liquid nitrogen. However, they also investigated Sn and Pb single crystals and polycrystals of Zn, Cd, Sn, Pb and In; also some tests were conducted at room temperature as well as 78K. They employed direct current (d.c.) pulses of the order of $10^3 A/mm^2$ in strength and $\sim$100μs duration by discharging a bank of capacitors. The time between pulses was of the order of 2-10s, which time was sufficient to allow the specimen to return to its original temperature following the current pulse. Most tests were performed in uniaxial tension on specimens 1mm dia. x 15mm long; however, some were conducted in compression on specimens 2.5mm x 6mm, respectively, to evaluate the role of thermal expansion due to Joule heating on the observed change in flow stress.

Single Crystals - An example of the early results obtained by Troitskii and coworkers on Zn single crystals is given in Fig. 1. The following

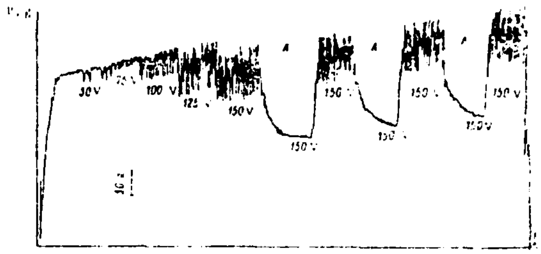

Fig. 1 Load versus extension diagram for Zn single crystals tested in uniaxial tension at 78K and a rate of $0.01cms^{-1}$ showing load drops due to the application of d.c. pulses produced by discharging capacitors with the voltages V indicated. Regions A correspond to stress relaxation (i.e., machine was shut off). From Troitskii and Rozno(4).

were noted: (a) the application of d.c. pulses during plastic flow at a constant extension rate produced load drops, whose magnitude increased with the capacitor voltage (i.e., with current density), and (b) the magnitude of the load drops was appreciably smaller during the later stages of stress relaxation compared to those during plastic flow at a constant extension rate. Furthermore, the load drop for a given voltage was larger for the tests at 78K compared to those at 300K. Also, there was a tendency for the load drop ΔP to increase as the level of the flow stress increased. In view of this, Troitskii and coworkers used the ratio $\Delta P/P$ as a measure of the electroplastic effect, where P is the total load (flow stress) prior to the application of a current pulse.

Figure 2 shows that the ratio $\Delta P/P$ varied with crystal orientation and with purity level. For 99.998% pure Zn crystals, the maximum value of $\Delta P/P$ ranged between $\sim$0.15-0.25 and occurred for χ_o = $\sim$60°-80° (χ_o is the angle between the

basal plane and the specimen axis), whereas for Zn crystals doped with 0.02% Cd the maximum value of $\Delta P/P$ was ~0.25-0.40 and occurred at $\chi_o \approx 45°$.

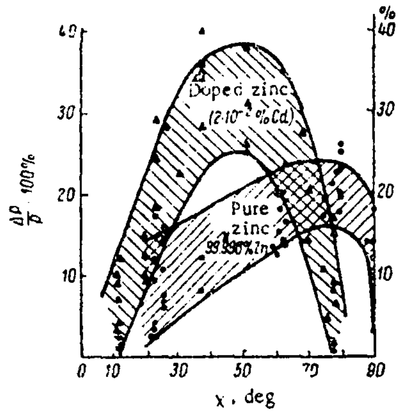

Fig. 2 The variation of $\Delta P/P$ with crystal orientation for pure Zn (99.998%) and Zn doped with 0.02% Cd for tests at 300K and a current density of ~1900A/mm². From Troitskii and Rozno(4).

In early studies(3,4) it was found that the magnitude of the load drop decreased with increase in strain rate. Later studies(8) however revealed that the effects of strain rate were not straightforward in that ΔP increased with strain rate at low rates, reached a maximum at an intermediate rate, and then decreased again at higher rates.

Changes in specimen length $\Delta \ell$ which occurred during a load drop were calculated through the relation $\Delta \ell = K\Delta P$, where K is the spring constant of the test system. The changes in length ranged from 2 to 18×10^{-3}mm for current densities from 600 to 1800A/mm²(10), which correspond to strains of $\sim 10^{-4}$ to 10^{-3} for the Zn single crystals employed. Examination of the crystal surface with a microscope revealed no significant change in the overall slip band structure(3,4), except perhaps some decrease in the coarse bands (which were separated 18-24μm) compared to thin bands (separated 9-12μm)(4).

Load drops were observed for compression as well as for tension(3,4), indicating that thermal expansion due to Joule heating was not the principal cause of the load drops resulting from the current pulses. Althoug a magnetic field of ~2000 Oe had no effect on the magnitude of the load drop, a change in polarity of the current pulse did(9,10).

Recognizing that such side effects of the current pulses as Joule heating, pinch effect and skin effect can contribute to the load drop, studies were carried out to evaluate the relative contribution of these side effects. Calculations and measurements(4,9,10) of the temperature rise ΔT due to Joule heating during a single current pulse of ~100μs applied to a specimen in liquid nitrogen yielded values of ΔT = ~0.4C for a current density J of 500A/mm² and 3-8C for J = 2000A/mm². Similar considerations for tests at room temperature with forced cooling yielded a maximum temperature rise of 12-16C. The thermal expansion produced by these temperature rises was estimated to be responsible for 3-4% of the load drop at 78K and 10-15% of that at 300K(4). The fact that load drops occurred in compression tests(3,4) supports the conclusion that thermal expansion was not responsible for a large fraction of the total load drop. Further support that thermal expansion cannot account for a large portion of the load drop in uniaxial tension is that the increase in specimen length (and in turn drop in load) produced by a current pulse increased by a factor of ~2.5 as χ_o increased from 18° to 44°(9).

The second influence of Joule heating is related to the effect of temperature on the flow stress, i.e., thermal softening. The fractional decrease in load expected due to this cause for the temperature rises given above was estimated (9,10) to be no more than 0.4-0.6% of the applied load, which represents at most only 3% of the total load drop ΔP. Thus, the combined effects of thermal expansion and thermal softening were estimated to account for only 6-7% of the magnitude of the observed load drops ΔP.

The magnitude of the skin effect was evaluated by placing thermocouples at the surface and in the interior of specimens and comparing the temperature at the locations produced by a current pulse. It was found that the difference in temperature at these positions for tests carried out in liquid nitrogen did not exceed 0.3C even at the highest currents, the higher value being in the interior. This was taken to indicate the absence of a skin effect. Worthy of note is that there was no difference in thermocouple readings between the center and ends of the specimen.

To measure the magnitude of the pinch effect, a two-part specimen configuration was employed (8,9). One part of the specimen (Part A) was deforming plastically under the applied load, whereas the other part (Part B) had no applied load. Current pulses were passed through each part separately and the associated load drops ΔP measured for pure Zn and Zn + 0.2% Cd alloy crystals. The results are presented in Fig. 3 The load drop ΔP_B associated with Part B, which measured only the pinch effect, was similar in magnitude for both materials and exhibited the same current density dependence. On the other hand, the load drop ΔP_A associated with Part A had a stronger dependence on current density than ΔP_B; moreover, the current dependence of ΔP_A was influenced by impurity content, being larger for the Zn + 0.2% Cd alloy. The difference $\Delta P_A - \Delta P_B$ was of the order of 75% of the total load drop ΔP_A for the Zn + 0.2% Cd alloy crystals and of the order of 50% for pure Zn, indicating that the pinch effect was responsible for ~25 to 50% of the total load drop in these crystals.

Since the combined effects of thermal expansion and thermal softening due to Joule heating were found to represent only a small fraction (6-7%) of the total load drop (ΔP_A above) resulting from a current pulse, Troitskii(3,4,9-11)

considered the difference $\Delta P_A-\Delta P_B$ to be due mainly to an interaction between drift electrons and dislocations. Support for such an interaction was provided by the observation that a threshold value of the current density (~400A/mm^2) was required for $\Delta P_A-\Delta P_B > 0$ and that a weak polarity effect (~10% of ΔP_A) occurred upon reversal of the current polarity, both effects being larger for the Zn + 0.2% Cd alloy.

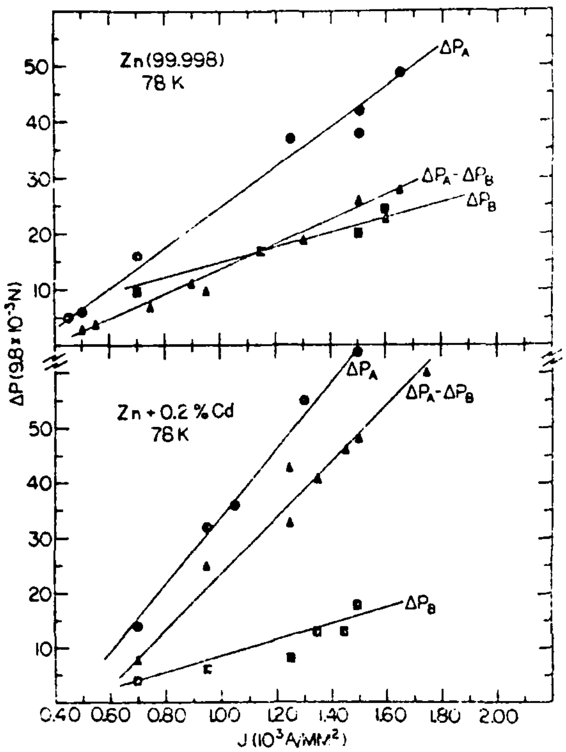

Fig. 3 ΔP_A (total load drop), ΔP_B (load drop due to pinch effect) and $\Delta P_A-\Delta P_B$ (load drop due to electron-dislocation interaction) versus current density for pure Zn (99.998%) and Zn + 0.2% Cd single crystals (χ_o = 18°) tested at 78K. Data from Troitskii(9). Spring constant of test system K = 3.37 10^{-2}mm/N.

Polycrystals - Troitskii and coworkers investigated the effect of single current pulses on polycrystals of Zn, Cd, Sn, Pb and In(3,4). The resulting load drops ΔP were similar in magnitude to those for single crystals tested under identical conditions. However, because of the larger values of the flow stress, the ratios $\Delta P/P$ for the polycrystals tended to be lower than those for single crystals. The values of $\Delta P/P$ at a condenser charge voltage V = 150 volts were 1.2% for Cd, 2% for Zn, 3% for Sn, 5% for Pb and 5% for In. These ratios are considerably smaller than the maximum values of 20-40% obtained for single crystals tested under similar conditions.

STRESS RELAXATION AND CREEP - The effects of d.c. current pulses on the stress relaxation of Zn, Cd and Pb crystals at 78K were studied by Troitskii and Stashenko(13) employing d.c. current pulses of 400A/mm^2 and 65μs duration at a frequency of 100Hz. These authors found that the rate of stress relaxation increased significantly with the application of the current pulses over that without pulses, the difference increasing with the stress level prior to the start of the relaxation. Moreover, a further increase in the rate and amount of stress relaxation occurred when the polarity of the current was reversed during the relaxation. The magnitude of the polarity effect was complex. It depended on the prior stress level and on the time following the beginning of the stress relaxation at which current was reversed, increasing with the prior stress level and decreasing with relaxation time prior to the current reversal. The relaxation rate did not rise immediately after the current was reversed, but only occurred after a certain lapse in time, which increased the longer the prior relaxation time before current reversal.

Subsequent studies(65,66) revealed that the enhancement of stress relaxation increased with increase in pulse duration t_p in the range 50-190μs for J = 200-250A/mm^2 and ν = 100Hz. Further, it was found that the influence of a change in polarity decreased with t_p and with a reduction in the time between the reversals.

Creep tests were conducted on Zn crystals at 78K employing d.c. pulses of 200-250A/mm^2 and 30-240μs duration at a frequency of 100Hz (16,17). In the first series of experiments(16) it was found that the electron-dislocation contribution to the thermal component τ^* of the creep stress decreased from approximately 80% to 10% as the pulse duration time increased from 60 to 150μs. In the second series of experiments (17) it was established that the creep rate increased significantly upon application of the d.c. current pulses, the magnitude of the increase being appreciably greater than that obtained from an a.c. current which produced an equivalent temperature rise. Contrary to the results from the first series of experiments, the magnitude of the increase in creep rate in the second series was enhanced with increase in pulse duration, the greatest effect occurring for durations of 110 to 150μs.

More recently Stashenko and Troitskii(67) investigated the effect of current (150-300A/mm^2) pulse frequency ν in the range of 2 to 600Hz on the creep of Zn and dilute Zn-Cd alloys at 78K. They found that the creep rate increased with frequency, the effect of ν being greatest between 2 and 100Hz. Analysis of the logarithmic creep curves indicated that the current pulses produced a change in the thermal component of the stress $\Delta\tau^*$ and in the strain hardening rate h, the former increasing and the latter decreasing with ν in the range considered.

FRACTURE AND FATIGUE - The effects of d.c. pulses of ~10A/mm^2 on the ductility of uncoated Zn crystals tested at 78K and on Zn crystals coated with a Hg film tested at 300K were investigated by Troitskii, Skobtsov and Men'shikh(21). They found that the current pulses increased the

resolved shear strain at brittle fracture by 100-120% at 78K and by 50-60% at 300K. Associated with this increase in fracture strain was a reduction in critical resolved shear stress and in the strain hardening rate. The beneficial effects of the current pulses increased with pulse duration from 5 to 150μs and with increase in frequency of the pulsing from 0.005 to $0.4s^{-1}$ (time between pulses of 200 and 2.5s, respectively). The load drop associated with a single pulse increased significantly for an increase in time between pulses from 2.5 to 20s, with only little further change between 20 and 200s.

In a later study, Troitskii(68) investigated the effects of a 5um thick Hg coating on the electroplastic effect in Zn crystals tested in tension at 78K with current pulses of $\sim 10^3 A/mm^2$, 100μs and 0.1Hz. It was found that the Hg coating increased the load drop due to the current pulses by 30 to 50%.

Golovin, Finkel and Sletkov(22) investigated the effects of d.c. pulses of the order of $10^3 A/mm^2$ 100μs on crack propagation in silicon iron sheet in tension. They found that both current density J and the time of application of the current t_a in relation to the delay time t_d for crack initiation without current (560μs) influenced crack growth. For a current density of $10^3 A/mm^2$ and $-200<t_a<0\mu s$, the initial crack hardly grew at all and was the most favorable time with regard to preventing crack growth. No retarding effect of current pulses on crack growth occurred if the current was applied prior to crack initiation ($t_a<-500\mu s$) or once crack growth was well on its way through the specimen width ($t_a>100\mu s$). In the latter case, the current pulse accelerated crack growth and melting occurred at the crack edges. The optimum current density for retarding further growth of a crack which had attained a length of 0.4 of the specimen width was in the range of $0.2\times10^3<J<4\times10^3 A/mm^2$; less retardation occurred for both lower and higher current densities.

A distinguishing feature associated with the retardation of the crack growth was the existence of a crater at the tip of the halted crack, namely an elliptical hole having an axis ratio of 1.2 to 1.5, with the major axis along the crack growth direction. The size of the crater varied from 10 to over 100μm depending on the pulse parameters. The edges of the craters showed signs of melting, indicating that the temperature rise at the crack tip rose to at least 1500C.

Karpenko et al.(23) investigated the effects of a low density ($0.07A/mm^2$) continuous d.c. current on the low-cycle fatigue of steel specimens in environments of air, H_2 gas and a 3% NaCl solution. The passage of the electric current increased the fatigue life in all three environments, the greatest effect occurring for tests in H_2. Metallographic examination revealed that slip bands were more uniformly distributed and the morphology of the fracture surface more uniform for specimens with current compared to those without.

In another study Troitskii(68) found that the application of high density direct current pulses ($\sim 10^3 A/mm^2$ 100μs and 0.2Hz) during cyclic loading (3 cycles) of Zn crystals up to 0.8 yield stress at 78K increased the subsequent yield stress by as much as 33% and the shear strain at fracture as much as 75% depending on crystal orientation.

DISLOCATION MOBILITY - Zuev et al.(18) investigated the effects of d.c. current pulses up to $200A/mm^2$ and 200μs duration on the mobility of $\{11\bar{2}2\}\langle 11\bar{2}3\rangle$ dislocations in Zn at 77 to 300K. The current was transmitted through the specimen normal to the $(11\bar{2}0)$ plane, both with the motion of the electrons along and counter the movement of the dislocations. An increase in dislocation velocity occurred for both directions of the current, the magnitude of the effect increasing with the current density but tending to saturate at $J>75A/mm^2$. The magnitude of the effect was greater when the motion of electrons coincided with that of the dislocations compared to when it was counter.

As a critical test for an electron-dislocation interaction, Boiko, Geguzin and Klinchuk(19) considered the direction of current flow on the contact area produced by the compression of a single crystal Cu sphere between two parallel Cu plates while undergoing d.c. pulses up to 30A with duration of $10^{-2}s$. Due to the smallness of the ball-plane contact area, current densities of $\sim 10^5 A/mm^2$ were obtained. The contract area was significantly larger at the plate where the direction of electrons enhanced dislocation motion compared to where it inhibited it. Further, the difference in size of the contact area increased with increase in current density.

OTHER EFFECTS - Troitskii and Linke(68) studied the emission of electrons during the plastic deformation of Pb, Cd, Zn, and In single crystals with and without the application of d.c. pulses of $\sim 10^3 A/mm^2$ and ~100us duration. Electron emission from the specimen increased significantly as they were deforming plastically. Additional bursts of emission occurred with the current pulses, the magnitude of the bursts increasing with current density and with pulse duration.

METAL WORKING - A review of the application of high density d.c. current pulses ($\sim 10^3 A/mm^2$ for ~100μs) to metal working is given in Ref. 32. The current pulses reduced the force required to draw Cu, stainless steel and W wires by as much as 30% and improved the postdrawing tensile properties. For Cu and W the effect increased linearly with current density and with the frequency of the current pulse for a polarity with plus upstream of the die so that the motion of electrons coincides with the direction of the deformation zone. Reversing the polarity also resulted in a decrease in the drawing force, but the magnitude was only about one-fourth that for plus upstream. Although a continuous d.c. current of equal average current density also lowered the drawing force and exhibited a polarity effect, the magnitude of the changes were not as large as those for the pulsed currents. The current pulsing also produced changes in the texture of the Cu wire and reduced the percentage of deformation-

induced α-phase in the stainless steel.

AMERICAN WORK

FLOW STRESS - One of the present authors (H.C.) and his coworkers(34-38,40,69) have investigated the effects of single, high density d.c. pulses ($\sim 10^3 A/mm^2$ for $\sim 50 \mu s$ duration) on the flow stress of a number of polycrystalline metals (Cu, Al, Pb, Sn, Fe and Ti) representing a range in crystal structures, valencies and stacking fault energies. A principal objective of the work has been to separate any electron-dislocation interaction from the side effects of the current pulses. One procedure employed to remove essentially all side effects except that due to thermal softening is illustrated in Fig. 4a. A single current pulse is applied during the plastic deformation of a wire specimen in uniaxial tension and the drop in stress $\Delta\sigma_p$ measured, which can amount to as much as 35% of the total flow stress, Fig. 4b. The specimen is then decrementally unloaded to the long range internal stress and the drop in stress $\Delta\sigma_E$ resulting from the application of the same stress pulse measured. $\Delta\sigma_E$ then includes such side effects as thermal expansion due to Joule heating, the skin effect and the pinch effect. The difference $\Delta\sigma_p$-$\Delta\sigma_E$ (Fig. 4c) should then consist primarily of thermal softening and any electron-dislocation interaction which may have occurred. For Ti the quantity ($\Delta\sigma_p$-$\Delta\sigma_E$) was found to increase with current density, but was relatively independent of test temperature between 77 and 300K and strain rate between 0.7×10^{-4} to $17 \times 10^{-4} s^{-1}$; it however increased with interstitial solute content in the range of 0.2 to 1.0 at.%(36).

Because of test machine inertial effects, one cannot measure the load drop during the short time ($\sim 50 \mu s$) the current pulse acts. Therefore, computer modeling was employed(37,38) to separate

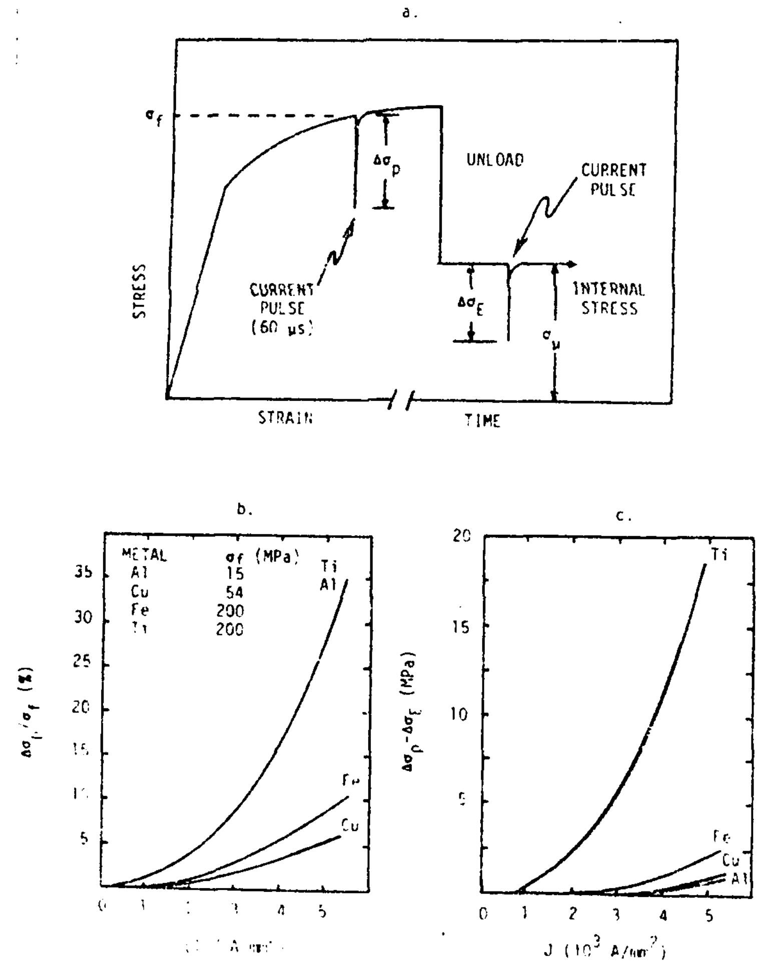

Fig. 4 Electroplastic measurements by the authors. (a) Experimental procedure, (b) Ratio of stress drop $\Delta\sigma_p$ to the flow stress σ_f for Al, Cu, Fe and Ti as a function of current density J, and (c) Difference in stress drop during plastic flow $\Delta\sigma_p$ and that at the long-range internal stress $\Delta\sigma_E$ as a function of J.

thermal softening from any electron-dislocation interaction. An example of the results of such modeling for Ti is given in Fig. 5, using an experimentally-derived, thermal-activation constitutive equation for the relationship between flow stress, temperature and strain rate and reasonable values for heat loss at the specimen surface. Such calculations indicated that a contribution in addition to thermal expansion, thermal softening, skin effect and pinch effect was needed to account for the observed drop in the flow stress. This contribution $\Delta\sigma_{ep}$ ($\Delta\sigma_{ed}$ in present terminology) was taken to be proportional to the current density, giving for Ti

$$\Delta\sigma_{ed} = 0.019\ J \quad (\mathrm{MPa}) \tag{1}$$

where J is the current density in A/mm^2. Although standard slow-response recorders were employed in the early measurements(37,38) of the stress drops and temperature changes, more recent work(40) employing high speed recording systems supports the validity of the results of the earlier measurements.

Further support that a significant portion of the flow stress drop which occurs with a current pulse is due to some factor other than Joule heating or a pinch effect is provided by recent results(69) on the effect of plastic strain ε on the quantity ($\Delta\sigma_p-\Delta\sigma_E$). It was found that ($\Delta\sigma_p-\Delta\sigma_E$) decreases with ε for Cu and Al but is relatively independent of ε for Fe and Ti (Fig. 6) in spite of the fact that Joule heating increases slightly with strain in each case due to the increased resistivity. As will be shown below, these effects of strain are in accord with what one might expect if the rate controlling mechanism was that commonly held to occur, i.e., intersection of forest dislocations for the FCC metals, overcoming some intrinsic Peierls-Nabarro type obstacle for Fe, and overcoming of interstitial solute atom obstacles for Ti.

Varma and Cornwell(41) investigated the effects of current pulses of 250-370A/mm^2 for several seconds on the flow stress of single and polycrystals of Al. The ratio of stress drop $\Delta\sigma_p$ to total stress σ_f ranged between 2 to 6% for polycrystalline specimens and 2 to 50% for single crystals, increasing with increase in voltage across the capacitor bank (current density) and with decrease in crosshead speed in the range of 10^{-4} to $10^{-5}s^{-1}$. The effects of crosshead speed became especially large in single crystals for strain rates below about $3x10^{-5}s^{-1}$.

Goldman, Montowidlo and Galligan(44) studied the effects of switching on and off d.c. current (2-8A/mm^2) and of d.c. pulses (1.4-8.0A/mm^2) on the flow stress of Pb crystals tested at 4.2K. At low current densities ($J \lesssim 3$A/mm^2) the specimen was superconducting and no effect of current on the flow stress was detected. At higher current densities the specimen became normal and a drop in stress occurred, which the authors concluded was entirely due to Joule heating.

BRAZILIAN WORK

Silveira and coworkers(46,47) investigated the effects of small continuous d.c. and a.c. currents (1.6A/mm^2) on the stress relaxation of Cu and Al. Both a.c. and d.c. currents were found to increase the relaxation rate at temperatures near 0.5 T_m; however, the effect was greater for the d.c. The effect decreased as the number of times the relaxation was carried out. Further, they found that the d.c. altered the dislocation arrangement in the Cu specimens in that there occurred a partial destruction of the cell structure even in the first relaxation cycle(49). This destruction became more pronounced as the number of relaxation cycles was increased. No such effects on dislocation structure were observed for Al.

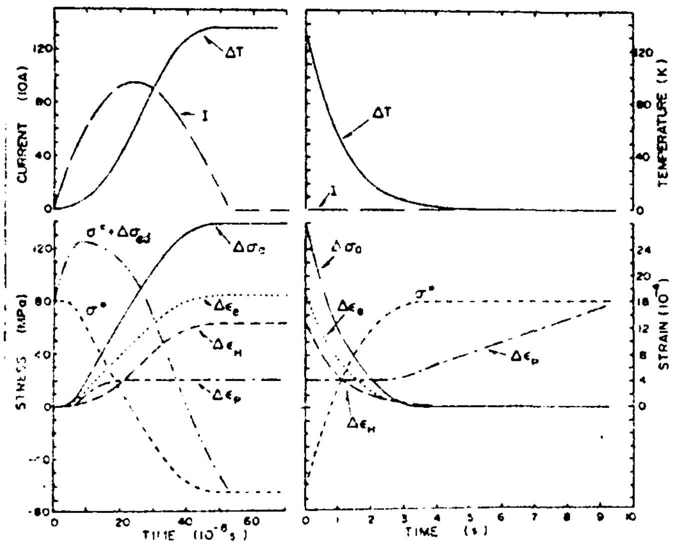

Fig. 5 Computer simulation of the increase in temperature ΔT, the thermal component of the flow stress σ^*, the stress due to an electron-dislocation interaction $\Delta\sigma_{ep}$, the drop in the applied flow stress $\Delta\sigma_a$, the corresponding elastic strain $\Delta\varepsilon_e(=\Delta\sigma_a/E_M)$, the strain $\Delta\varepsilon_H$ due to thermal expansion and $\Delta\varepsilon_p(=\Delta\varepsilon_e-\Delta\varepsilon_M)$ the plastic strain due to $\sigma^* + \Delta\sigma_{ep}$ for polycrystalline zone refined Ti at 300K and $\dot{\varepsilon} = 1.7 \times 10^{-4}s^{-1}$ for a current pulse I ($J = 5000$A/mm^2), assuming $\Delta\sigma_{ep} = 0.019J$. From Okazaki, Kagawa and Conrad(37).

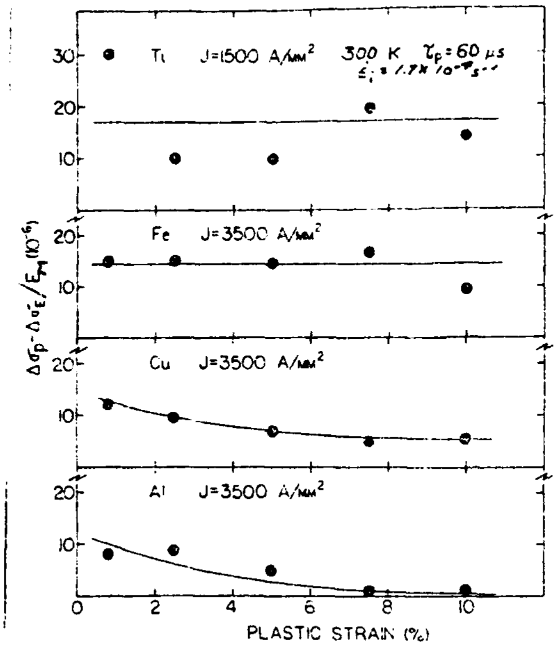

Fig. 6 Effect of plastic strain on $(\Delta\sigma_p - \Delta\sigma_E)/E_M$ for a current pulse of $\sim 10^3 A/mm^2$ and 60 μs applied during the uniaxial tension of polycrystalline Al, Cu, Fe and Ti at 300K and $\dot\varepsilon = 1.7 \times 10^{-4} s^{-1}$. E_M=machine-specimen modulus.

EAST GERMAN WORK

San Martin and coworkers(45) studied the effects of heating V_3Si single crystals by continuous d.c. and a.c. current ($\sim 25 A/mm^2$) on creep behavior at 1450-2000K. Electric current heating increased the steady state creep rate over that observed for indirect heating, with d.c. heating being more effective than a.c. During high temperature deformation of V_3Si, Kramer(50) found for indirect heating of the specimen that the slip system {100}<001> was active with straight dislocations of edge orientation. For specimens heated with d.c. current, no slip plane was favored. Since the creep of V_3Si appears to be controlled by dislocation climb, San Martin et al.(45) propose that the enhanced dislocation motion by the electric current might be due to electrontransport of point defects.

DISCUSSION

The studies reviewed above into the effects of electric current (i.e., drift electrons) on the flow stress or creep of metals can be classified into two general categories: (a) the effects of high density d.c. pulses ($\sim 10^3 A/mm^2$ for 10^{-5} to 10^{-4} s) at relatively low temperatures, i.e., $T/T_m < 0.4$ and (b) the effects of relatively low density ($1\text{-}10 A/mm^2$) continuous d.c. (or a.c.) current at $T > 0.4T_m$. Some of the results which are strongly supportive of the idea that the influence of the high density current pulses reflects the interaction of drift electrons with crystal defects in general or dislocation motion in particular include: (a) the effect depends on crystal orientation or relative direction of electron motion compared to dislocation motion, (b) the effect depends on the current polarity and (c) it decreases with strain in FCC metals. Support for drift electron-crystal defect interactions being responsible for enhanced stress relaxation and creep rates at high temperatures includes: (a) the observed changes in slip and dislocation structure and (b) no significant side effects are expected for the test conditions that were employed. The discussion to follow will focus on the mechanism(s) whereby high density current pulses influence the plastic flow at low temperatures, since most of the work of the present authors, and others, has been in this area.

Theoretical calculations of the force exerted by drift electrons on dislocations in a number of metals are presented in Table 1. It is here seen that the force varies by about 2 orders of magnitude depending on the theory considered.

To analyze the effect of drift electrons on the plastic flow of metals, it is here assumed that the plastic flow is governed by a thermally-activated rate equation of the form

$$\dot\varepsilon = \dot\varepsilon_o \exp - \{\Delta G(\sigma^*)/kT\} \qquad (2)$$

where $\dot\varepsilon$ is the plastic strain rate, $\dot\varepsilon_o$ the pre-exponential factor and ΔG the Gibbs free energy of activation, which is a decreasing function of the effective stress $\sigma^* = \sigma - \sigma_\mu$. σ is the applied stress and σ_μ is the long-range internal stress. For small changes in σ^*, Eq. (2) can be expanded to give for the strain rate prior to the application of a current pulse

$$\ln\dot\varepsilon_i = \ln\dot\varepsilon_o - \frac{\Delta G^*}{kT} + \frac{v\sigma^*}{kT} \qquad (3)$$

where $v = -\partial\Delta G/\partial\sigma^* = kT\partial\ln\dot\varepsilon/\partial\sigma$ is termed the activation volume. Upon application of a current pulse we have

$$\ln\dot\varepsilon_{ed} = \ln\dot\varepsilon_{o_{ed}} - \frac{\Delta G^*_{ed}}{kT} + \frac{v(\sigma^* + \sigma^*_{ed})}{kT} \qquad (4)$$

where σ^*_{ed} is the additional stress acting on dislocations due to the force exerted by the drift electrons. The subscript ed refers to the values associated with the current pulse. Subtracting Eq. (3) from Eq. (4) gives

$$\ln(\dot\varepsilon_{ed}/\dot\varepsilon_i) = \ln(\dot\varepsilon_{o_{ed}}/\dot\varepsilon_o) - \frac{(\Delta G^*_{ed} - \Delta G^*)}{kT} + \frac{v}{kT}\sigma^*_{ed} \qquad (5)$$

According to the theoretical equations in Table 1, σ^*_{ed} is proportional to J. Eq. (5) then indicates that for a constant current density, $\ln\dot\varepsilon_{ed}$ should

Table 1 - Theoretical estimates of the force per unit length (f/ℓ) exerted on dislocations by drift electrons.

Metal	b (10^{-8}cm)	$n^{(1)}$ (10^{22}e/cm^3)	$E_F^{(2)}$ (ev)	$v_F^{(3)}$ (10^8cm/s)	$(f/\ell)/J$ $\left(\frac{\text{dyne/cm}}{\text{A/cm}^2}\right)$ Eq. (1) (10^{-10})	Eq. (2) (10^{-}[illegible])	Eq. (3) (10^{-}[illegible])
Ti	2.95	0.38	0.886	0.56	22.80	[illegible]	[illegible]
Cu	2.55	8.33	6.94	1.56	1.04	8.5[illegible]	2.6[illegible]
Al	2.86	6.25	5.73	1.42	1.38	8.[illegible]	[illegible]
Fe	2.48	5.83	5.47	1.39	1.48	7.9[illegible]	2.6[illegible]

Theoretical Equations:

Kravchenko(5): $f/\ell = [\frac{b}{4} (\frac{3n}{2E_F} \frac{\cdot^2}{v_F})](v_e - v_d) = (\frac{3b}{8} \frac{\cdot^2}{eE_F v_F})J$ [illegible] (1)

Klimov, Shnyrev & Novikov(6): $f/\ell = \frac{1}{3} n\, m^* b\, v_F\, (v_e - v_d) = (\frac{m^* b\, v_F}{e})J$ (2)

Roshchupkin, Miloshenko & Kalinin(7): $f/\ell = 4\hbar n\, (v_e - v_d) = (\frac{4\hbar}{e})J$ (3)

Electron velocity: $v_e = J/en$ Dislocation velocity: $v_d \approx 0$ J = current density

NOTES: (1) n = electron density, from R. A. Brown, J. Phys. F: Metal Phys. 7 [illegible]

(2) E_F = Fermi energy = $\frac{\hbar^2}{2m_e}(3\pi^2)^{2/3} n^{2/3}$; $\hbar = \frac{6.624\times10^{-27}}{2\pi}$ erg-sec = [illegible]

(3) v_F = Fermi velocity = $(2E_F/m_e)^{\frac{1}{2}}$

vary as v, providing $\dot{\varepsilon}_{oed}$ and ΔG^*_{ed} do not change with the factors which produce a change in v. For FCC metals v decreases with strain ε, whereas for BCC metals and Ti, v is independent of strain (70,71). Hence $\dot{\varepsilon}_{ed}$ is expected to decrease with ε for FCC metals, and to remain relatively independent of ε for BCC metals and Ti, which was experimentally observed (Fig. 6). $\dot{\varepsilon}_{ed}$ is here considered to be equal to $(\Delta\sigma_p - \Delta\sigma_E)/E_M t_p$, where t_p is the pulse duration (60μs).

Assuming that $\dot{\varepsilon}_{oed}$ and ΔG^*_{ed} do not vary with strain, one can on the basis of Eq. (5) determine σ^*_{ed} for FCC metals from the effect of strain on v and $\dot{\varepsilon}_{ed}$. Results obtained(69) for the application of high density d.c. pulses (2500-5500A/mm^2 for 60μs) during the plastic flow of polycrystalline Cu in uniaxial tension at 300K were employed in this analysis. The maximum temperature rise in these tests was 5K. Again $\dot{\varepsilon}_{ed}$ was taken to be $(\Delta\sigma_p - \Delta\sigma_E)/E_M t_p$.

Since for FCC metals $v = kT/\sigma m$(70), where $m = \partial\ln\sigma/\partial\ln\dot{\varepsilon}$ is relatively independent of ε, Eq. (5) becomes for this class of metals

$$\ln(\dot{\varepsilon}_{ed}/\dot{\varepsilon}_i) = [\ln(\dot{\varepsilon}_{oed}/\dot{\varepsilon}_o) - (\frac{\Delta G^*_{ed} - \Delta G^*}{kT})] + \frac{\sigma^*_{ed}}{m}(\frac{1}{\sigma}) \quad (6)$$

Hence a plot of $\ln(\dot{\varepsilon}_{ed}/\dot{\varepsilon}_i)$ versus $1/\sigma$ for a given current density J should yield a straight line of slope σ^*_{ed}/m. Data for Cu are plotted in this manner in Fig. 7, yielding reasonably straight lines whose intercepts and slopes increase with J. The values of σ^*_{ed} were determined from the slopes of the lines by taking $m = 5\times10^{-3}$(70) and plotted versus J in Fig. 8. The slope of the straight line passing through the origin [illegible] $\sigma^*_{ed}/J = 8.06$(dyne/cm^2)/(A/cm^2). The force per unit length (f/ℓ) on the dislocation is given [illegible]

$$(f/\ell)/J = \sigma^*_{ed} b/MJ \quad (7)$$

where M is the Taylor orientation factor. Substituting $\sigma^*_{ed}/J = 8.06$(dyne/cm^2)/(A/cm^2), b = 2.55×10^{-8}cm and M = 3.18(72), one obtains [illegible] = 6.5×10^{-8}(dyne/cm)/(A/cm^2), which is in reasonable accord with the value of 2.6×10^{-8}(dyne/cm)/(A/cm^2) calculated employing the equation of Roshchupkin et al.(7) in Table 1.

The intercepts in Fig. 7 yield the quantity $[\ln(\dot{\varepsilon}_{oed}/\dot{\varepsilon}_o) - (\Delta G^*_{ed} - \Delta G^*)/kT]$. A log-log plot of this quantity versus J is given in Fig. 9,[+] yielding a straight line of slope of 2. This suggests that this quantity varies as $(\sigma^*_{ed})^2$. Assuming that the drift electrons have only little effect on the obstacle strength, i.e., on ΔG^*, the intercepts would then reflect the effect of electrons on the pre-exponential $\dot{\varepsilon}_o = \rho_m A b \nu^*/\ell^*$, where ρ_m is the mobile dislocation density, A is the area swept out per successful thermal fluctuation and ν^* the frequency of vibration of the dislocation segment of length ℓ^* between obstacles. The second power of σ^*_{ed} would [illegible] accord with an effect of σ^*_{ed} on the mobile dislocation density ρ_m(73). To be noted from Fig. [illegible] is that $\dot{\varepsilon}_{oed}/\dot{\varepsilon}_o = 1$ when J = 600A/mm^2, suggesting a threshold value of J (or σ^*_{ed}) for its effect on ρ_m.

[+] In Fig. 9 it is assumed that $(\Delta G^*_{ed} - \Delta G^*)/kT = 0$.

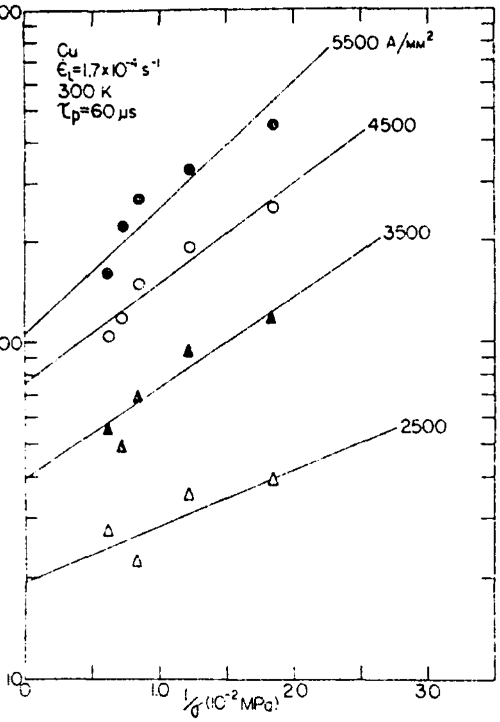

Fig. 7 $\ln(\dot{\varepsilon}_{ed}/\dot{\varepsilon}_i)$ versus $1/\sigma$ as a function of current density for polycrystalline copper.

Since in the case of Fe and Ti, v does not vary with ε, one cannot separate the effects of electrons on σ^* from their effects on ΔG and $\dot{\varepsilon}_o$ in the manner of FCC metals. The order of magnitude greater value of $(\Delta\sigma_p-\Delta\sigma_E)/E_m$ for Ti (Fig.4) compared to the FCC and BCC metals could be due to a significantly larger effect of drift electron on ΔG, i.e., on the strength of the interstitial solute obstacles, which are the rate controlling obstacles in this metal(71).

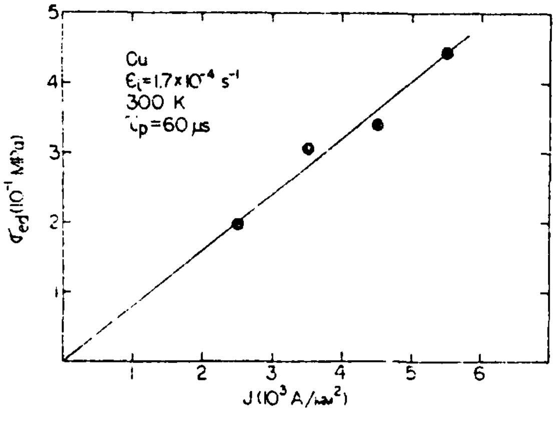

Fig. 8 σ^*_{ed} determined from the slopes of the straight lines in Fig. 7 versus current density.

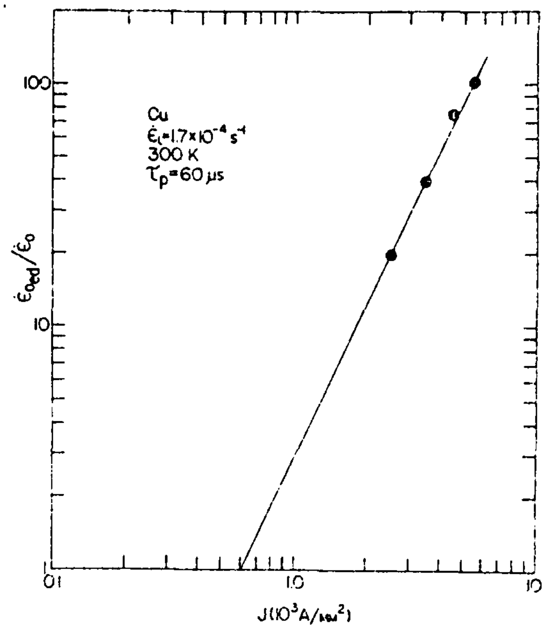

Fig. 9 $\ln(\dot{\varepsilon}_{oed}/\dot{\varepsilon}_o)$ determined from the intercepts of Fig. 7 versus current density. It is here assumed that $(\Delta G^*_{ed} - \Delta G^*)/kT = 0$.

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